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DEFECTS IN INGOTS

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13. ABSTRACT

The document is a translation of Chapter VII of the book entitled "Ingots of Titanium Alloys" and discusses defects in such alloys under three categories: a) defects introduced by the charging material; b) defects which are formed during melting and in preparation for melting; and c) defects of shrinkage origin. Methods of preventing and eliminating defects of each category are presented. These include proper selection of materials, removal of shavings, improved technical procedures in melting, the use of ultrasonic flow detection and the limitation of flame cutting where it is possible.

Details of illustrations in this document may be better studied on microfiche.

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CHAPTER VII

DEFECTS IN INGOTS

V. I. Dobatkin, N. F. Anoshkin, et al.

The technology of the production of titanium alloys ensures obtaining an ingot without gas pores, sharply defined liquation, shrinkage porosity, concentric and radial cracks (defects characteristic for steel ingots and ingots of non-ferrous metals). Nevertheless, in ingots of titanium alloys defects are encountered which can be subdivided into three types:

a) defects introduced by the charging material;

b) defects which are formed during melting and in preparation for melting;

c) defects of shrinkage origin (intracrystalline fractures).

DEFECTS INTRODUCED BY CHARGING MATERIAL

Sponge

Into the charge of commercial titanium alloys usually more than 60% titanium sponge is introduced. This predetermines its

main role in forming the quality of the ingot (the level of mechanical properties, the content of harmful alloying elements and contamination of the ingot by impurities.

Various adverse factors (excessively "hot" conduct of the process with magnesium thermal reduction, a high content of impurities in argon, magnesium, and titanium tetrachloride, inleakage during distillation, etc.) lower the quality of sponge titanium. Furthermore, if there is no careful checking on the dressing of the bloom and the classification of the sponge, then commercial batches are contaminated with small pieces containing an increased quantity of gas and metallic impurities. The most frequently encountered defective small pieces of sponge can be divided into four types (Tables 27, 28).

Table 27. Chemical composition of defective sponge.

(1) Тип дефектной губки	(2) Химический состав, %							
	Mn	Fe	Si	Mg	Cr	Al	O_2	N_2
1	0,014	0,013	0,02	5,8—0,09	0,03	0,04	3,9	0,025
2	0,064	3,1	0,12	0,04	0,19	0,69	0,6	0,067
3	0,05	0,02	0,05	0,01	—	0,06	0,3	0,022
4	0,07	3,7—17,0	0,15	—	4,81	0,03	0,2	0,011

KEY: (1) Type of defective sponge. (2) Chemical composition.

Table 28. Mechanical properties of defective sponge.

(1) Тип дефектной губки	Механические свойства (2)			
	$\sigma_в$		δ %	ψ %
	kg/mm^2	mN/m^2		
1*	665	(67,9)	—	—
2	770	(78,7)	21,7	39,3
2	770	(78,7)	24,2	46,8

* Хрупкое разрушение. (3)

KEY: (1) Type of defective sponge. (2) Mechanical properties. (3) Brittle rupture.

1) sponge with a high content of magnesium and magnesium chloride;

2) dark sponge (usually obtained from the "bottoms" of the bloom);

3) sponge which is oxidized (with temper color) and burned;

4) sponge containing an increased quantity of iron, silicon, and chromium (usually obtained from the "sides" of the bloom).

Table 27 gives the composition of the defective small pieces selected from several batches of sponge, and Table 28 gives the mechanical properties of the first two types of defective sponge.

The tensile strength of sponge according to certificate comprised 422-490 MN/m^2 (43-50 kg/mm^2).

A considerable content of gas impurities in the defective sponge, a high melting point of titanium oxides and nitrides, and their great density as compared with the melt lead to the fact that the inclusions are preserved in the ingots despite double remelting with expendable electrodes. The lower the grade of sponge the greater, as a rule, the number of defective pieces. In ingots of alloys VT1-2, VT3-1, and 3Al, for the melting of which sponge of the lowest grades was used, very frequently inclusions of a gray color were observed containing up to 2.2% O_2 and up to 0.8% N_2 (Fig. 96). The microhardness of such inclusions is within the limits of 6.4-10.8 H/m^2 (650-1100 kg/mm^2), exceeding by two or three times the microhardness of the parent metal.

Artificial contamination of a charge of alloy OT4 by defective small pieces of all four types led to the appearance of sections enriched by oxygen (up to 0.24%) in the sheets. The structure of such sections is distinguished by clearly expressed banding and by the presence of cracks (Fig. 97).

Fig. 96. Oxygen-enriched inclusion in an ingot of 350 mm diameter of alloy VT3-1 observed at a depth of 130 mm from pouring face, ×3.

Fig. 97. Microcrack in a sheet of alloy OT4 contaminated by small pieces of oxidized sponge, ×30.

Sponge of high grades is more uniform in composition; however, even it is frequently not free from defective small pieces. In recent years the quality of sponge has continuously improved. Nevertheless, the further perfection of all procedures in the production of sponge, the most rigid classification checking, and rejection of sponge by the manufacturers and input control at the metallurgical plants are indispensable conditions for obtaining high-grade ingots devoid of internal defects. In this respect,

the creation recently of standards for defective sponge which include all four types of contamination must be considered as a great step forward.

Shavings

Titanium ingots are turned, cut, and faced with cutters with hard-faced soldered bits made of alloy VK8 (92% WC, 8% Co), having a density of 0.015 Mkg/m^3 (14.7 g/cm^3) and a melting point of 3000°C.

During the processing of ingots under the effect of thermal and impact stresses, the hard-faced bits gradually disintegrate and their particles remain as shaving fragments. The harder and tougher the alloy being worked the more shaving fragments fall (Fig. 98).

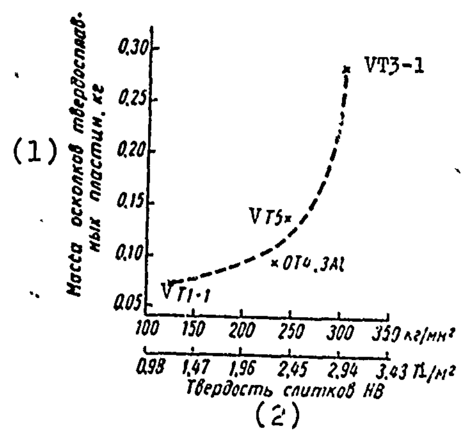

Fig. 98. The quantity of fragments of hard-faced bits of alloy VK8 entering, per 1000 kg of shavings in the machining of ingots of various titanium alloys. KEY: (1) Mass of fragments of hard-faced bits. (2) HB hardness of the ingots.

In a charge of titanium alloys up to 20% ground shavings is used. In grinding the shavings and in pressing the expendable electrodes, the bits falling as shavings are crushed, forming in the ingots the whole series of inclusions in sizes from 1 to 20 mm. Such inclusions were discovered during X-raying of a number of sheets of one of the titanium alloys (Figs. 99, 100).

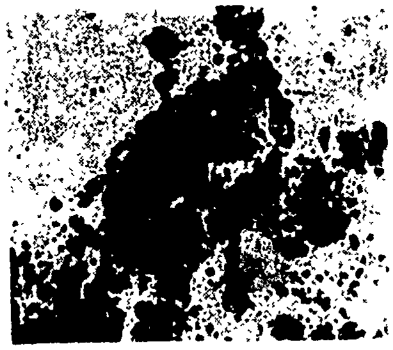

Fig. 99. X-ray photograph of a section of sheet with inclusions of fragments of hard faced tips.

Fig. 100. An inclusion of hard-faced tip in an ingot of alloy VT1-2, ×2.

The fragments of hard-faced tips can be eliminated from shavings with the aid of a magnetic separator.

Numerous experiments in the artificial introduction into the shavings of fragments of tips of alloy VK8, and also the experience in series production ingots, into the charge of which they introduced up to 30% shavings, showed the high reliability of the separator ensuring in practice complete extraction of all fragments found in the shavings. The exception amounted to only those small (a few mg) fragments which were welded to the shavings in the mechanical treatment of the ingots (Fig. 101).

Experiments in the artificial contamination of pressed electrodes with such fragments showed that the latter are dissolved during two remelts and do not give rise to the formation of inclusions in ingots and sheets exposed to X-raying.

Fig. 101. Shaving with a fragment of hard-faced tip welded to it. Sizes of fragment 0.45 × 5 × × 10 mm, mass 0.46 μg, (0.46 g).

Master Alloy

Of all the master alloys utilized to obtain titanium alloys, the most refractory, and therefore the most dangerous in the sense of formation of inclusions, is the alloy Al-Mo.

In ingots of alloys VT14, VT8 and VT3-1 frequently fine inclusions of metallic molybdenum and brittle gas-saturated impregnations were encountered. In order to remove them, for alloys VT8, VT14, and VT3-1 they now use master alloys with 40-60% Mo having a melting point of 1100-1300°C; they leave a "bog" on the bottom of the crucible during the pouring of the master alloy from the ladle, containing small pieces of unmelted molybdenum; they pour the master alloy from the ladle over the cells of the mold in a vacuum, which precludes the formation of oxides and prevents them from falling into the master alloy. An optimum composition of the master alloy Al-V may be considered a composition with a vanadium content within limits of 60-80% (melting point 1600-1750°C). A necessary condition when using alloy Al-V is the careful removal of the crust of slag from the surface of the bloom. Otherwise, inclusions of calcium oxide will remain in the ingots.

DEFECTS WHICH ARE FORMED DURING THE WELDING OF INGOTS

Inclusions Enriched by the Alloying Components and by Gaseous Impurities

In thick sheets of some titanium alloys containing aluminum there was discovered a seam with dispersed brittle flattened out

inclusions of a brown or gray color (Fig. 102). Such inclusions are distinguished by high aluminum content (7-20%) and in most cases by a high content of nitrogen (1.4-13.5%) and oxygen (0.6-3.0%). Sound metal contains about 4% aluminum, less than 0.5% nitrogen, and less than 0.2% oxygen.

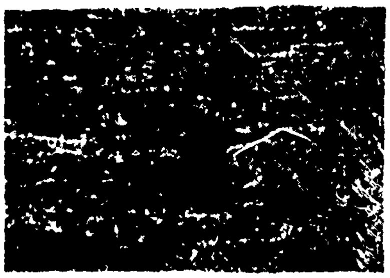

Fig. 102. A seam in a sheet of titanium alloy.

In ultrasonic flaw detection of ingots, inclusions of such composition did not come to light, nevertheless, some specific features of vacuum melting make it possible to assume a foundry character of origin of the inclusions mentioned.

In industrial furnaces, ingots melt under a pressure (in the chamber and the vacuum conductor) of approximately 39.9-79.8 N/m^2 ($3\text{-}6 \cdot 10^{-1}$ mm Hg). In accordance with the calculations and experiments of a number of researchers, pressure in the arc combustion zone should be on an order higher, i.e., 133-798 N/m^2 (1-6 mm Hg). Taking into account that during melting the tip of an expendable electrode 60-100 mm in length is heated to 1200-1500°C, and the superheating of the surface of the fluid bath under the electrode reaches 300°C (see Fig. 24), it is possible to expect that during the first and second remelts there can be metal not only from the expendable electrode. This can be:

a) condensate and reguluses of solidified metal from the walls and flanges of the crystallizer and chamber walls, of cinder of the adapter and the lateral surface of the expendable electrode – as a result of vibration and jolting of the furnace induced by the working of the vacuum pumps and by pronounced changes in the combustion behavior of the arc (by short circuits);

b) the corona of the built-up ingot together with the sublimates deposited on it – as a result of the premelting of the base of the corona;

c) the corona of the ingot of the first remelt and splashes in the zone of welding – at the end of melting before changing to the condition of removal of the pipe;

d) the built-up edges of semiliquid metal by the bead surrounding the end of the expendable electrode or adapter and being formed under the effect of electrodynamic forces and erosion of the cathode in the process of removal of the pipe by a lower-power arc.

After unloading the ingot of the first remelt from the furnace, a coating of condensate remains on the surface of the ingot in the form of a two-layered crust 0.5-2 mm thick (Figs. 103, 104).

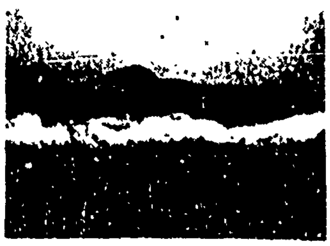

Fig. 103. Crust of condensate on the surface of an ingot: the white layer – magnesium chloride; the dark layer – magnesium, aluminum, solidified drops of titanium, ×5.

Fig. 104. The structure of the surface of the dark layer in the condensate crust, ×5.

The external (bright) layer consists basically of magnesium chloride and magnesium; the internal, having a metallic fracture, consists of a condensate of aluminum, magnesium, and minute droplets of the melt. Below, the composition is given (in %) of the sublimates, corona, splashes, and the internal part of the crust (the layer) in the melting of ingots with a diameter of 650-850 mm of alloy VT5 (I – first remelt; II – second remelt):

	Алюминий (1)	Азот (2)
Конденсат на стенках кристаллизатора: (3)		
I	16,6—26,2	0,02—0,163
II	4,6—6,7	0,05—0,09
Конденсат на стенках камеры: (4)		
I	12,6	0,05
II	—	—
Конденсат на огарке: (5)		
I	13,3—19,5	0,6
II	4,0—5,1	—
Корона слитка, I (6)	7,8—12,5	0,06—0,126
Пленка, I (7)	17,9—11,4	—
Основной металл, II (8)	4,6	0,03

KEY: (1) Aluminum; (2) Nitrogen; (3) Condensate on walls of crystallizer; (4) Condensate on chamber walls; (5) Condensate on the cinder; (6) Corona of ingot; (7) Film; (8) Basic metal.

In the melting of ingots of the second remelt, the process of sputtering of the melt and sublimation from the lateral surface of the ingots of the first remelt as well as from the surface of the bath is repeated, although to a lesser degree.

By themselves, sublimates still are not reflected in the quality of the ingot. However, the vaporized metal is a distinct getter; it absorbs the air which gets into the furnace through leaks in the vacuumed volume of the furnace. Furthermore, in the premature de-vacuuming of the furnace after the melting of the ingots or welding of the expendable electrodes one to another and to the adapter, the condensate having an exceptionally developed surface, interacts intensely with nitrogen and oxygen. These two factors create a real possibility of contamination of the ingot by inclusions. In fact, the melting points of aluminum

trioxide (2050°C), of magnesium oxide (2800°C), chromium oxide (2245°C), stannic oxide (1960°C), titanium oxide (2020°C), titanium nitride (3205°C) are significantly higher than the melting point of the titanium alloys. The probability of the formation of the inclusions increases if sponge of the lowest grade is used in which the magnesium content is relatively great.

With a large inleakage into the furnace during welding or in the premature de-vacuuming of the furnace the leakages are oxidized and their surface has a temper color from yellow to blue. The content of oxygen in the leakages varies within limits of 0.11-0.5%; of nitrogen 0.025-0.29%, of hydrogen 0.009-0.011%.

During the melting of the electrode in the zone of welding, the cross connection which connects the leakages with the electrode may premelt, and the semi-solid leakages will fall into the melt.

In a similar way, the built-up edges behave which are formed on the tip of the expendable electrode at the moment of removal of the pipe by a low-power arc. They are more dangerous than the leakages because they are immersed in the melt, from which superheating is already almost completely removed. In Fig. 105 in the axial part of the hole, in its bottom, there are visible dark spots of elongated shape. In their origin they are due to the built-up edges which were formed on the adapter made of alloy of 3Al during the removal of the pipe, and then they fell into the melt.

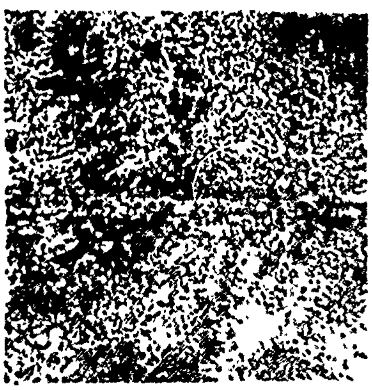

Fig. 105. Macrostructure of an ingot of 850 mm diameter alloyed with aluminum (I. V. Polin).

The dropping in of the built-up edges which are formed during the removal of the pipe can be eliminated in two ways: in the first place, ingots intended for production of sheets should be melted without removing the pipe; in the second place, in setting up the conditions for removal of the pipe the period during which the forming built-up edges fall off should be reduced.

The following ways to remove the inclusions which are formed during the melting of ingots may be proposed:

1) more rigid norms of permissible inleakage before melting;

2) sufficiently complete cooling of ingots of the first remelt after melting and of the expendable electrodes after welding in a non-depressurized furnace, which prevents oxidation of the built-up edges;

3) washing and cleaning of the ingots of the first remelt for the purpose of removal of the condensate layer not welded with the ingot;

4) removal of the coating of condensate from the internal surface of the crystallizer, chamber, cinder, adapter, etc., after each melting;

5) obligatory removal of the corona before welding on of the ingots of the first remelt;

6) machining of the oxidized surface of the ingots of the first remelt;

7) obtaining ingots of the first remelt with well-melted surface (without cold laps, cavities, etc.) and with a closely welded on layer of condensate.

Brittle Gas-Saturated Inclusions of Products of Flame Cutting

The production of titanium ingots in vacuum electric arc furnaces causes the necessity for cutting of the ingots: the removal of the corona of the ingots of the first remelt before their welding to one another in the furnace of the second remelt, the removal of deposits and splashes in the zone of welding, separation of poorly welded together expendable electrodes. In all the cases enumerated they usually resort to the use of flame cutting by an oxyacetylene flame. In the cutting zone elevated temperatures develop (on the order of 3000°C) as a result of the exothermicity of the combustion processes of acetylene and the interaction of titanium with oxygen and nitrogen, and also with the combustion products. As a result, a different type of gamma titanium compound is formed with nitrogen, oxygen, carbon and hydrogen.

Some properties of the compounds being formed and of solid solutions are noted below [39, 43, 55, 59]:

Тип соединения (1)	TiO	Ti_2O_3	TiO_2	TiC	TiN	Твердый раствор в титане (2) N_2	Твердый раствор в титане (2) O_2
Область гомогенности (3), % (ат.)	48–54	59–62	—	18–50	30–50	0,10	0,33
Температура плавления (4), °C	1750	2130	1825	3147	3205	1940	1900
Плотность (5): kg/m^3	4930	4600	4180–4250	4980	5430	—	—
g/cm^3	4,93	4,6	4,18–4,25	4,98	5,43	—	—

KEY: (1) Type of compound; (2) Solid solution in titanium; (3) Region of homogeneity; (4) Melting point; (5) Density.

In the metal which was subjected to flame cutting, three zones may be noted which differ in the degree of contamination by harmful impurities: porous slag of yellowish-brown color, a

brittle layer having a temper color with finely crystalline fracture, and gas-saturated metal.

The chemical composition of the products of the flame cutting of ingots of the alloy OT4 is as follows, % (by mass):

	O_2	N_2	H_2	C	Al	Mn
Шлак (1)	4,8—1,16	1,26	0,02	0,16	0,53	0,81
Хрупкий слой (2)	3,7—0,72	1,49	0,034	1,11	2,1	1,13
Газонасыщенный слой (3)	0,3	0,02	0,01	0,05	2,3	1,2
Основной металл (4)	0,15	0,04	0,01	0,05	2,3—2,6	1,3

KEY: (1) Slag; (2) Brittle layer; (3) Gas-saturated layer; (4) Basic metal.

The slag and the brittle layer are depleted of alloying components (aluminum and manganese) as a result of their evaporation and burn up. The density of the slag (determined by the water displacement method) is equal to 4650 kg/m^3 (4.65 g/cm^3), of the brittle layer 4950 kg/m^3 (4.95 g/cm^3).

The microstructure of the zones of the effect of flame cutting is shown in Fig. 106.

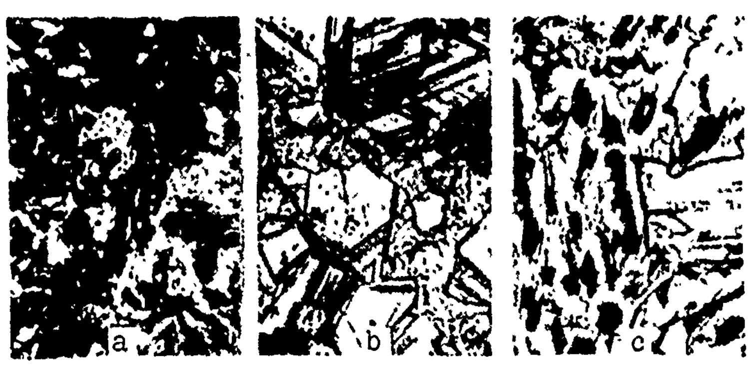

Fig. 106. Microstructure of zones of the effect of flame cutting, ×200: a) slag; b) brittle layer; (c) gas-saturated layer.

The microhardness of the slag, the brittle layer, the gas-saturated layer, and the parent metal is equal to 14.7; 4.9, and 2.16-2.74 H/m^2 (1502, 500 and 220-280 kg/mm^2), respectively.

The size of the gas-saturated layer is approximately 0.25 mm.

The high melting points of the compounds and the solid solutions which represent component parts of the slag and the brittle layer, and also the significant density of the latter (as compared with the density of the melt permit supposing a low probability of their melting or dissolving when they fall into the molten metal.

Actually, the experiments conducted in artificial contamination of ingots of 350 mm diameter of the alloy OT4 by the products of flame cutting by means of notching the ingots of the first remelt by an oxyacetylene flame confirmed the possibility of the formation of brittle inclusions in the ingots of the second remelt, leading to appearance of stratified layers in sheets (Fig. 107).

Fig. 107. Stratification in sheets, induced by slag and brittle inclusions dropping into the ingot: a) fracture of sheets of alloy OT4, 13 mm thick, ×1; b) microstructure of inclusions, ×340.

Microhardness in the stratified zone comprises 4.1-5.3 H/m^2 (420-540 kg/mm^2), the oxygen and nitrogen content in the inclusions reaches 0.34 and 0.21%, respectively.

All three flame cutting zones can cause the formation of inclusions in ingots and semi-finished products. While separately introduced in the expendable pressed electrodes, the small pieces of gas saturated metal representing each of the zones, led to the

formation of inclusions in the ingots and further – to the appearance of inclusions in the sheets, stratified layers, cracks, and sections contaminated by gas impurities, with coarse, badly deformed structure.

In the light of what has been expounded, the use of flame cutting in the production of titanium ingots is extremely undesirable and should be reduced to the minimum. Machine cutting of coronas, the pronounced rise in intensity of the magnetic field of the solenoid in furnace welding, the use of argon-arc welding of ingots of the first remelt – all these measures make it possible to almost completely eliminate flame cutting in the casting house. In cases when nevertheless it is necessary to resort to it, the cutting zone should be eliminated by machine treatment or by other means.

INTRACRYSTALLINE FISSURES

Intracrystalline fissures were first discovered in fractures of templates cut from ingots of alloys VT5-1 and 3Al with a diameter of 350 mm. For fractures of commercial medium-alloyed alloys a ductile transcrystalline fracture is characteristic; in this case the cross-section has a coarse grain and a laminar character of texture.

In the structure of the fractures of mentioned templates of alloys VT-5 and 3Al sharply delineated bright planes are visible commensurable in size with diameter of the grains (Fig. 108).

Fig. 108. Macrofracture of an ingot of 350 mm diameter of alloy 3Al.

A similar structure is frequently encountered in the bottom part of big ingots of alloys of 3Al and 2Al in the fracture of the boss after removal of the notched bottom template, and also in the fracture at the place of welding of the ingots of the first remelt to one another. The simplest means of detection of structural defects in an ingot is ultrasonic flaw detection. In sounding ingots with intracrystalline fissures, the oscillogram is usually characterized by several impulses which testify to the presence of sources of reflection of ultrasonic vibrations [UZK] (УЗК) located at different depths (Fig. 109).

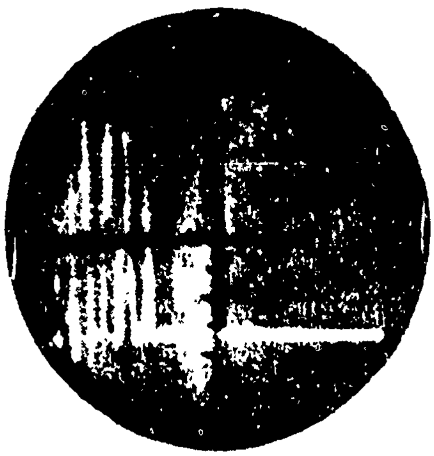

Fig. 109. Impulses on the scanner of the flaw detector corresponding to sources of reflection of UZK located at different depths of an ingot.

With the movement of the probe over the plane surface of the ingot the earlier appearing impulses disappear, and new ones appear which differ in intensity and place of location on the scanner. It is possible to approximately mark the area (coordinates) of the existence of intracrystalline fissures: in a plane perpendicular to the axis of the ingot, the outline being delineated by the probe on the end of the ingot inside which in sounding there appear defect impulses on the oscillograph; in the diametric plane (axial section) – a distance between minimum and maximum depth of the occurrence of the sources of UZK reflection (the depth of occurrence is fixed by the depth gauge of the instrument). If from the volume thus found of the ingot cross template is cut, faced, and then etched, then on the surface of a large section it is possible to see fine fissures. These fissures

are commensurable in length with the grain size and, as a rule, are limited by dimensions of the grains. If we notch the template and then break it, then in the fracture it is always possible to see bright planes, whereby these planes in a cross-section view of one half of the template is a mirror image of the planes in the cross-section view of its other half. The emergence of these planes onto the plane of the section are also marked by fissures (Fig. 110).

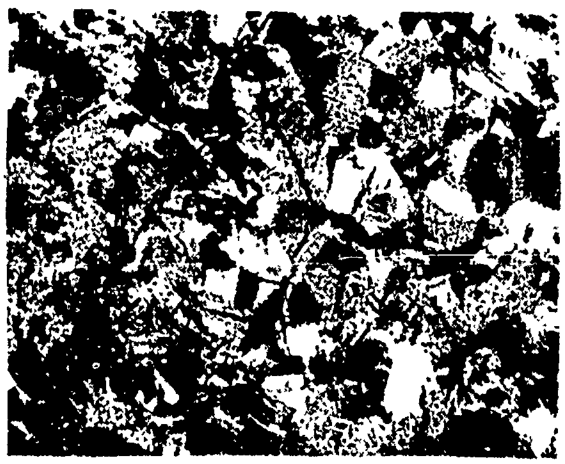

Fig. 110. Macrostructure of cross section of an ingot of 350 mm diameter of alloy VT5 in the zone of intra-crystalline fissures.

Similar phenomena in studying the template can be encountered even in the case when the defect on the oscillograph looked like a single one (i.e., on the scanner of the instrument only one impulse was recorded.

In a cross section of the ingot the intracrystalline fissures are arranged, as a rule, in the central region, and in the zone of equiaxial crystallites, preferably in the bottom part. Prior to the present time, intracrystalline fissures have been discovered in ingots of alloys 2Al, 3Al, VT5, VT5-1, OT4-2, alloyed with more than 2.5% aluminum. In commercially pure titanium and two-phase alloys with a large content β-stabilizers fissures were

not encountered (the exception is alloy VT8). Depending on the dimensions of the grain, the fissures have a length of from 3-4 up to 40 mm, and under visual inspection they seem to be macrofissures, but unlike hot and cold radial and concentric fissures which are characteristic for nonferrous and ferrous ingots [14, 24, 44] in their essence are microcracks because they are arranged within the limits of a grain. Depending on the composition and the diameter of the ingots structural damage occurs in from 30-60% of the ingots of the alloys mentioned above.

Research on Intracrystalline Fissures

The microstructure of cast metal of α-alloys is characterized by the presence within the limits of one crystallite (or converted β-grain) of several blocks with different directions of α-plates, whereby within the limits of one block the direction of the α-plates is identical (see Chap. VI).

As a rule, fissures are limited by the dimensions of one block and they pass at an angle to the α-phase plates. However, cases of crack propagation to an adjacent block or even adjacent grain are frequent (Fig. 111). The bright planes along which the metal fails during its fracture, within the limits one grain go in steps parallel to one another; on microsections very frequently it is possible to observe within the limits of one block several usually parallel fissures. This makes it possible to assume that the metal fails along specific crystallographic planes. Actually, γ-ray definition of the indices of planes along which the cracking occurs in ingots of alloys VT5 and 3Al confirmed that in all cases this was the basal plane of the hexagonal close-packed lattice (0001) which for metals with such a lattice, including cast titanium alloyed with aluminum, and having large grain dimensions, is one of the basic (if not only) shear and cleavage planes [16, 20, 39, 43].

Fig. 111. Microstructure of an ingot of 350 mm diameter of alloy 3Al with fissures located within the limits of one block, ×200.

In examining the microstructure of ingots (Fig. 111) it is evident that the fissures intersect the α-plates whose direction up to and beyond the fissure in all cases remains constant. Moreover, in some sections of the microsections it is distinctly evident that part of the α-plates remained on one side of the fissure, and part – on the other with exceptionally precise matching of the longitudinal boundaries of the plates. Therefore the fissures appeared even after the formation of the α-plates; i.e., below the temperature of allotropic transformation. Otherwise, if the fissures arose at the temperature of the β-region or at the initial moment of phase change β → α, then they would become the lining at phase recrystallization and the coincidence of direction of the α-plates up to and beyond the fissure would be absolutely non-obligatory.

In order to clarify the nature of fissures, an examination was made of the content of alloying component (aluminum) and impurities (iron, silicon, carbon, hydrogen, nitrogen) in approximately five thousands ingots of alloys 3Al, VT5-1, and VT5 of various sizes. The statistical treatment of these data made it possible to ascertain that in the ingots of any one alloy the

formation of fissures was not connected with an increase in the content of alloying components or impurities. It is only possible to note that for individual alloys and individual dimensions of ingots, the average content of hydrogen in ingots with fissures is more than in those without them. Furthermore, this difference is expressed predominantly in ten thousandths of a percent.

The determination of the chemical composition of metal in the zone of fissures in alloys 3Al and VT5-1 makes it possible to consider with confidence that the emergence of fissures in ingots in practice is not connected with liquation in macrovolumes and contamination of ingots (also in macrovolumes) by impurities of iron, silicon, carbon, nitrogen, and oxygen.

The microhardness of ingots of 350-850 mm diameter in the zone of sound metal as well as in zones where intracrystalline fissures have been discovered is identical in practice. The values of microhardness measured from the cleavage planes are exceptionally unstable and fluctuate for alloy 3Al within limits of 2.8-15.2 H/m^2 (290-1600 kg/mm^2) exceeding on the average by 588 MN/m^2 (60 kg/mm^2) the value of microhardness in the sound and defective zones of the ingot.

If during the melting of ingots through leaks in the vacuum-tight seal of the tray air gets into the furnace, then the nitrogen and oxygen contained in it are completely absorbed by the surface layers of the ingot heated to elevated temperatures, by the corona and by the rim of the fluid bath. If the leak is great or the air gets in through the seals of the ingot mold, rod, etc., then on the surface of the bath there is formed a fine oxide film which by the pressure of the arc is constantly driven off to the walls of the crystallizer. Therefore, frequently the ingot which was melting in the presence of significant inleakage has a solid, nitrogen and oxygen saturated crust (the thickness of which depends upon the intensity of the inleakage) and a rather plastic center

whose composition in respect to gas impurities does not differ from an ingot melted under optimum conditions.

Hydrogen behaves completely differently. Possessing exceptionally high diffusion mobility and much greater solubility in β-phase (up to 50% at) as compared with α-phase (up to 7% at) [37, 39, 43], hydrogen as the surface layers of the ingot cool and phase change β → α flows in them hydrogen diffuses in the internal, more heated layers of the ingot in which the process of phase change lags as compared with the surface layers. Therefore, the internal layers of ingot, in which the formation of intracrystalline fissures is observed, can be somewhat enriched by hydrogen as compared with the peripheral sections of the ingot, in so doing, the enrichment can bear a selective character, extending preferably to specific planes of the close-packed hexagonal lattice [37].

Actually, the hydrogen content determined by the spectral method in an ingot of 850 mm diameter of alloy 2Al, on the cleavage planes (the basal planes) was found within limits of 0.008-0.020% (on the average 0.013%) while in the zone of columnar crystals as well as in the zone of fissures (but not on planes) its average content comprised 0.005%.

The results of analyses had a definite error because on the discharge surface there were small cracks and inequalities. Nevertheless, it can be noted that on planes on which disruption of the continuity of metal occurs, the hydrogen content is greater than in any other place on the section of the ingot.

Electron-microscope investigation of the cleavage planes in ingots of alloy 3Al with magnification of 17,000 times made it possible to discover on their surface octahedrons of rather regular form with the size of diagonal on the order of 1 μ (Fig. 112). ' e discovered octahedrons may be separations of particles of any phase or be the result of condensation of vacancies. In the latter case, the faces of the octahedrons have been

formed by packing defects. A similar case linked with the formation of tetrahedrons (the faces of which consist of packing defects on the slip planes in gold) is examined by V. A. Pavlov [46]. The segregation of impurities in the form of a finest layer on the basal planes confirms the fact that after a brief etching of the planes microhardness increased and octahedrons were not observed on them.

Fig. 112. Structure of cleavage planes in an ingot of 350 mm diameter of alloy 3Al, ×17,000

Dependence of Crack Formation Upon Magnitude and Character of Stress Distribution in an Ingot

In Chapter IV the question of magnitude and character of shrinkage stresses in the ingots of titanium alloys was examined. The connection between stresses and intracrystalline fissures can be proven by the following reasoning:

1. <u>The content of aluminum in ingots.</u> As was noted earlier, with an increase in the content of aluminum in an alloy, tensile stresses increase in the central part of the cross section of the ingot, which is explained by the increase in the heat resistance of the alloy and correspondingly an earlier (i.e., at elevated

temperatures) change of the metal of ingot into the region of elastic deformations. The treatment of statistical data in the ultrasonic flaw detection of ingots of 350-450 mm diameter of alloys VT5, VT5-1, and VT8 shows that with an increase in the aluminum content in the alloy there is also an increase in the number of ingots affected by intracrystalline fissures (Fig. 113)

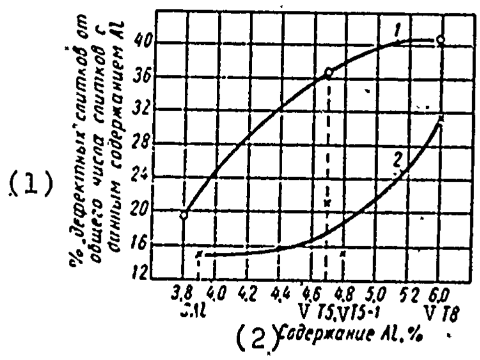

Fig. 113. Dependence of the number of defective ingots upon the aluminum content of alloys: 1 – diameter of ingots 350 mm; 2 – diameter of ingots 450 mm. KEY: (1) % of defective ingots from the total number of ingots with the given content of Al. (2) Al content, %.

It must be noted that with an increase in the aluminum content in the alloy along with the growth in strength and heat resistance the plasticity of ingot in the cast state is lowered which cannot intensify the process of crack formation.

2. <u>The distribution of intracrystalline fissures vertically in an ingot.</u> In the direction from the sprue to the bottom in the cross section of ingots there is noticed a regular increase (approximately twofold) in tensile stresses. In the same direction there is an increase in the frequency of the development of intracrystalline fissures with the aid of ultrasonic flaw-detection equipment (Table 29). With the exception of ingots of 350 mm diameter, the number of ingots with fissures (sources of UZK reflection) in the bottom part is one and a half to two times more than the number of ingots with fissures in the pouring part.

3. <u>The diameter of ingot.</u> Tensile stresses in the cross section of ingots increase with the diameter of the ingot. Thus, for alloy 3Al stresses in the bottom and pouring parts of an ingot

Table 29. Statistical data on the development of fissures (zones of reflection of UZK) in ingots.

(2) Сплав	(3) Диаметр слитка мм	(1) Число зон отражения УЗК					
		(4) в слитке		(5) в донной части слитка		(6) в литниковой части слитка	
		(7) абс.	%	абс.	%	абс.	%
3Al	350	370	100	108	29,1	262	70,9
	450	110	100	70	63,6	40	36,4
	650	33	100	20	60,6	13	39,4
	750	53	100	36	67,9	17	32,1
	850	103	100	82	79,6	21	20,4
BT5	350	267	100	91	34,1	176	65,9
BT5-1	350	174	100	51	29,3	123	70,7
	450	8	100	4	50,0	4	50,0

KEY: (1) Number of zones of UZK reflection; (2) Alloy; (3) Diameter of ingot; (4) In ingot; (5) In bottom part of ingot; (6) In pouring part of ingot; (7) Abs.

of 850 mm diameter by one and a half times exceed the corresponding stresses in an ingot of 350 mm diameter. In the same direction and approximately in the same relationship there is a change in the number of ingots having intracrystalline fissures (revealed by ultrasonic detection [UZD] (УЗД), depending on the size of the diameter of the ingots (Fig. 114).

(1)

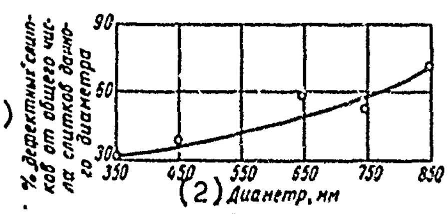

Fig. 114. Dependence of number of defective ingots upon the diameter of ingot. KEY: (1) Defective ingots from the total number of ingots of given diameter; (2) Diameter.

The presented regularities speak persuasively about the direct connection between tensile stresses of shrinkage origin in ingots and the process of crack formation.

But are sufficiently great stresses being developed in ingots to lead to the breaking of metal on the basal planes of a hexagonal lattice? As already mentioned in Chapter IV, the tensile stresses in the central part of the cross section of ingots reach 255-294 MN/m^2 (26-30 kg/mm^2) for alloys 2Al and 3Al and approximately 588 MN/m^2 (60 kg/mm^2) for alloy VT5. These values amount to 50-70% of tensile strength and are approximately equal to the plastic limit. It must be noted that the stresses of the first kind determined by the radiographic method are averaged in a circular belt of given radius in the cross section of the ingot. The stresses in any determined directions can be great; furthermore, it is necessary to take into account the possible local additional increases in stresses as a result of the appearance of stress concentrators (the imperfection of the lattice structure, the segregation of impurities, grain boundaries, etc.).

In connection with the remarks made above, tensile stresses in favorably oriented planes can reach values which exceed the plastic limit and the yield point, and give rise to the achievement of a critical value of elastic lengthening and the appearance of cracks.

Dependence of Crack Formation Upon the Lattice Parameters

The appearance of fissures in ingots, as was noted above, is characteristic for α-alloys alloyed with more than 2.5% aluminum (by mass).

In ingots of two-phase alloys as well as in the commercially pure titanium, fissures have not been discovered. The cause of this, in our opinion, consists of the following.

Commercially pure titanium has a c/a ratio equal to 1.587; in the interval from 20-800°C it is possible to note three acting elements of the shear of planes (0001), (1010), and (1011). The direction of shear for all planes is identical: (1120) [43].

With the introduction of aluminum into the alloy, the c/a ratio increases to 1.598-1.601 at 3 and 6%, respectively [39], and the mechanism for deformation changes in the direction inherent in an ideal close-packed hexagonal lattice with a c/a ratio equal to 1.633. In this case, shear passes predominantly along the basal plane (0001). In two-phase α + β-alloys an additional three systems of elements of shear are added: (110), (112), and (123), inherent to a bodycentered cubic lattice [32].

Thus, the conditions for relaxation of stresses which appear during the cooling of ingots, for solid solution alloys alloyed with aluminum, are much worse than for commercially pure titanium and two-phase alloys.

Summing up what has been expounded earlier, let us briefly formulate our presentations regarding intracrystalline fissures.

Intracrystalline fissures are microcracks in the central sections of the section of ingots. They are formed as a result of brittle breakaway (spalling) on the basal planes under the action of shrinkage stresses. Breakaway is facilitated by the segregation of impurities on the basal planes and by raising the c/a ratio of the lattice parameters of single-phase α-alloys alloyed with aluminum.

Behavior of Fissures During Mechanical Working of Ingots

During research on structural defects not one case was noticed where a fissure traversed a large number of blocks or grains and actually became a macrofissure.

If in the cast state fissures localize themselves within the limits of blocks and grains and in this sense they do not present a danger to the integrity of the ingot, then will they not be origins of transcrystalline or intercrystalline fissures

during the mechanical working of the ingot? For the purpose of clarifying this question the following experiment was conducted. From a zone with intracrystalline fissures of an ingot of 850 mm diameter of alloy 3Al cylinders 60 mm in diameter and with a height of 200 mm were cut. After etching on the lateral and end surfaces of the cylinders, a large number of fissures were revealed (Fig. 115). The cylinders were pressed into bushings from alloys VT1-1 and 3Al, whereupon the ends of the bushings were sealed in a laboratory vacuum electric arc furnace. Since the buildup of the plugs was conducted in a vacuum, then between the cylinder, the bushing, and the plugs no air layers remain and the cylinder was located in a tightly adherent vacuumed shell.

Fig. 115. Macrostructure of the end surface of a cylinder with fissures.

Billets thus prepared at 1000-950°C were forged into rods of 50 mm diameter. From the rods longitudinal large sections in diametric plane were manufactured (Fig. 116).

As can be seen from the presented photographs of the large sections, the cylinder of alloy 3Al in drawing the shell of the

Fig. 116. Longitudinal large sections of rods pressed into a bushing: a) shell of alloy VT1-1; b) shell of alloy 3Al.

rod in the shell of alloy VT1-1 was partially cracked and crushed. It is possible to see spallings of the alloy 3Al sprinkled in the die of alloy VT1-1, and conversely, sections of alloy VT1-1 (more plastic than alloy 3Al) which penetrated through the fissures into the deforming cylinder of alloy 3Al. On the section there were discovered no stratifications, fissures, or any other form of discontinuity; i.e., during hot deformation setting (welding) of the metal of the bushing and cylinder occurred. A large section of the rod obtained as a result of forging of the cylinder pressed into the bushing of alloy 3Al has a completely different character. In view of the fact that the cylinder and bushing have been manufactured from one and the same alloy, crushing of the cylinder did not occur. Just as also in the first case, the macrostructure is characterized by the absence of any form of discontinuity, which testifies to the reliable setting of the material of the bushing and the cylinder and complete welding of cracks which existed in the cylinder.

Samples of cast metal with fissures coming through to the surface could not be forged without a shell because the metal was oxidized in the fissures and failed during forging.

The peripheral sections of the cross section of an ingot which are found in a state of compression do not have pores, fissures, and the other metallurgical flaws (besides cold lap) and are

uniquely durable and form a monolithic shell for the center of an ingot with intracrystalline fissures. Therefore, during plastic deformation of ingots, the fissures are sealed up, and in so doing, the physical properties of the metal in the zone of former structural defects do not differ from the properties of sound metal.

The question of the behavior of metal after plastic deformation with more complex and more specific methods of testing deserves further study.

Since intracrystalline fissures can be encountered not only in ingots, but also in the weld seams of large-size parts and since the behavior of metal after crack sealing has not been clarified by more complex and more specific testing methods than short term breaking, the development of measures for elimination of fissures is very essential.

As preliminary measures the following may be proposed:

a) the introduction of β-stabilizators into single-phase α-alloys alloyed by aluminum in a quantity sufficient for fixation in the cast metal of very small quantities of β-phase;

b) a reduction in harmful impurities (iron, carbon, silicon, nitrogen, oxygen) in the charging material;

c) a reduction in the hydrogen content in ingots by means of installing of high-vacuum pumps on the furnaces which ensure the creation of vacuum during melting on the order of 1.33 N/m^2 ($1 \cdot 10^{-2}$ mm Hg);

d) grinding of the cast grain, which will make it possible to increase the concentration of impurities on the basal planes.

Whereas the last three propositions require checking, the first one causes no doubt whatsoever because it is confirmed by much work experience with various alloys.